A ULTRAHIGH STRENGTH MEDIUM MN STEEL MANUFACTURED BY WARM ROLLING PROCESS*

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Abstract

A ultra-high strength medium Mn steel with good ductility was designed and manufactured via the combination of warm rolling and V alloying, which produced not only the extensive precipitation of fine VC particles in both ferrite and austenite but also the bimodal size distribution of retained austenite (RA) grains. A substantial increase of yield strength up to 650 MPa was achieved due to the addition of V without deteriorating ductility. The coarse RA grains firstly transformed to martensite during yielding at the much higher stress threshold, followed by a significant work hardening due to dislocations multiplication, twinning-induced-plasticity (TWIP) and transformation-induced-plasticity (TRIP) in the ultrafine RA grains. The best combination of strength and ductility included 1.5 GPa ultimate tensile strength and 28% total elongation. This was achieved due to the sustainable TRIP and TWIP effects in the later straining stage resulting from the ultrafine austenite grains, the latter were reversely transformed from the recrystallized ferrite grains.

Keywords: Medium Mn steels; Vanadium carbides; Transformation induced plasticity, Twinning induced plasticity.

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1 INTRODUCTION

The present research is to manufacture a medium Mn steel that exhibits ultrahigh strength and good ductility at the same time, i.e. 1.5 GPa ultimate tensile strength (UTS) and 15% total elongation (TE). To achieve this goal, we chose 0.3% C–10% Mn as the basis for the new compositional design because this particular composition led to 1.2 GPa UTS and 65% TE simultaneously [1-6]. Next, the precipitation strengthening strategy via the typical microalloying of V was adopted since such an excessive value of TE could afford a wide window on sacrificing ductility for the effective precipitation hardening. The precipitation strengthening increment of 300 MPa could be realized by the addition of 0.7% V into this composition according to the calculation using the classical Ashby–Orowan equation. Since the complete precipitation of 0.7% V as carbide should consume about 0.15% C, the C content of designed steel was set as 0.45%. Moreover, 2% Al was added to raise the ferrite-to-austenite transformation temperature for faster reverse transformation kinetics and suppress the precipitation of cementite. Therefore, the composition of studied steel was finalized as 0.45C–2Al–10Mn–0.7V (all in weight percentage unless stated elsewhere). In addition, the warm rolling process was employed for this V-alloyed medium Mn steel, rather than the conventional hot-rolling or cold-rolling processes. The achieved microstructures and mechanical properties of designed steel were then presented and discussed in this paper.

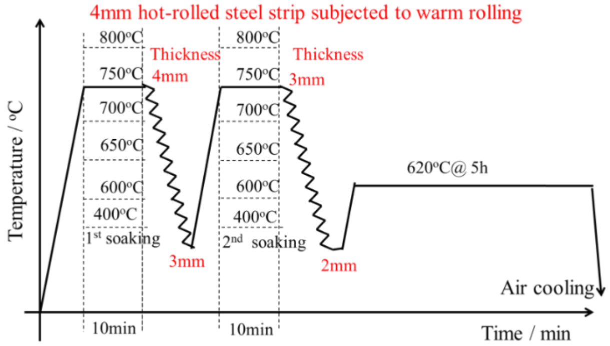


Figure 1. The sketch map of warm rolling processes.

2 MATERIAL AND METHODS

The two studied steels were melted in a 50 kg vacuum induction furnace and cast into ingots. One has the nominal composition of 0.47C-9.97Mn-0.7V-2.01Al; the other has the composition of 0.45C-9.71Mn-1.99Al, serving as the reference since it does not contain V. The compositions of the two samples are given in Table 1. The cast ingots were hot-forged into several 40 mm-thick billets, which were then heated to 1150 °C for 2.5 h for the solution treatment to fully dissolve the vanadium carbonitride. The billets were hot rolled to a thickness of 4 mm by six passes in a pilot two-high hot rolling mill with a finish temperature ~800 °C, followed by water cooling to room temperature to suppress the precipitation of VC. The 4-mm-thick hot rolled strips were further warm rolled to either 2 mm or 1.5 mm in thickness at temperatures in the range 250–800 °C in a four-high cold rolling mill. For the warm rolling, the hot rolled strip was first heated at the specified warm rolling temperature for 10 min, namely, the soaking treatment, and then rolled to achieve a total reduction of about 25% in three rolling passes. The process of soaking and warm rolling was repeated until the final thickness of 2 mm or 1.5 mm was obtained. The targeted thickness of 2 mm requires this process to be repeated two times at the temperature of 400–800 °C, and four times at 250 °C due to a higher resisting force at low warm rolling temperature. This process was repeated three times at 600–750 °C for the 1.5 mm-thick samples. A detailed description of warm rolling processes is given in Figure 1. The exact manufacturing process for each sample is listed in Table 1.

3 RESULTS AND DISCUSSION

Figures 2a and 2b reveal that the addition of vanadium clearly leads to a very significant increase in both UTS and yield strength (YS) by about 600 MPa without sacrificing much of TE. In parallel, higher warm-rolling reduction in thickness results in larger elongation and higher YS but has

a less influence on UTS for both the steels. In particular, a relatively larger Lüders strain is observed in the V-10Mn steel when compared to the 10 Mn steel.
A systematic examination results on the microstructures after the intercritical annealing at 620 °C for 5h on the studied samples at the different thickness along the transverse direction (TD) are shown in Figure 3. Ultrafine ferrite and austenite grains, having the equiaxed morphology and the size of several hundreds of nanometers, were observed in almost all the samples. They are mostly located in both the initial martensite bands and the intersections of slip lines, where martensite and ferrite were severely deformed during the previous warm rolling so that they recrystallized to ultrafine ferrite grains; next, some ultrafine ferrite grains reversely transformed to the austenite grains having the similar sizes during the intercritical annealing. The increase of warm rolling temperature from 400 °C to 600 °C leads to greater extent of recrystallization in ferrite/martensite. Nevertheless, the further increase to 750 °C produced much fewer ultrafine austenite grains but more coarse martensite, because almost all of the initial martensitic bands in the hot rolled structure transformed to austenite during the soaking at 750 °C.
It is shown in Figure 4 that many coarse austenite grains in V600H have transformed but the ultrafine ones still remain until the true strain of 9%. Before the deformation of V600H, the ultrafine austenite and ferrite grains appeared along the coarse austenite grain boundaries; the coarse austenite grains transformed during yielding and some of ultrafine austenite grains either transformed to martensite or twinned during the further straining until fracture. Figure 5 shows that the vanadium carbide particles with the sizes of tens of nanometers extensively precipitate in both ferrite and austenite after the intercritical annealing.
Figure 2 shows that the addition of V has led to the increase of YS of 10Mn steel by 500–650 MPa, which shall be firstly attributed to the pronounced precipitation of nanosized VC particles in both austenite and ferrite in the V-alloyed steel. This can be quantified using the well–known Ashby–Orowan equation [7]. Both the fraction and average size of VC precipitate were determined from the measurements. The calculated increments of YS are in the range of 310–370 MPa for both austenite and ferrite, which are much lower than the observed increments of 500-650 MPa. Therefore, there should be other factors can contribute to strengthening.
The microstructures in the samples of R600L, R600H, V600L and V600H have been examined in details and shown in Figure 6. It can be clearly seen that there are more both coarse-sized and ultrafine austenite grains in the V-alloyed steel than those in the V-free one. Therefore, the refinement of coarse austenite grains due to the V alloying should not be the reason for strengthening.
The VC nanosized precipitates could retard both recovery and recrystallization, leading to the decreased rate of dislocation annihilation [8, 9]. By using XRD, a much higher dislocation density in the austenite of 10Mn-V steel than that in 10Mn steel was indeed measured after the same process, see Figure 7. Therefore, it may be concluded that both precipitation and dislocation hardening should be mainly responsible for the significant increase of YS for the V-alloyed 10Mn steel.ng deformation, as discussed later.

4 CONCLUSIONS

The combination of 0.7% V alloying and warm rolling leads to a substantial strengthening up to 650 MPa for 10 Mn steel. This has produced not only the bimodal distribution of retained austenite grains but also the extensive precipitation of nanosized VC particles in both austenite and ferrite grains. It is concluded that precipitation strengthening and dislocation hardening, both resulting from the extensive precipitation of VC, are the main contributor to the increase of YS.

Table 1 Warm rolling processes employed in the present study (initial thickness=4 mm)

No.	Sample	Steel grade	Warm rolling process			
			Soaking temperature (°C)	Number of soaking cycles	Final thickness (mm)	Total reduction ratio
1	V250L	V-10Mn	250	4	2.0	50%
2	V400L	V-10Mn	400	2	2.0	50%
3	V600L	V-10Mn	600	2	2.0	50%
4	V650L	V-10Mn	650	2	2.0	50%
6	V750L	V-10Mn	750	2	2.0	50%
8	V600H	V-10Mn	600	3	1.5	63%
9	V650H	V-10Mn	650	3	1.5	63%
10	V750H	V-10Mn	750	3	1.5	63%
A	R600L	10Mn	600	2	2.0	50%
B	R650L	10Mn	650	2	2.0	50%
C	R600H	10Mn	600	3	1.5	63%
D	R650H	10Mn	650	3	1.5	63%

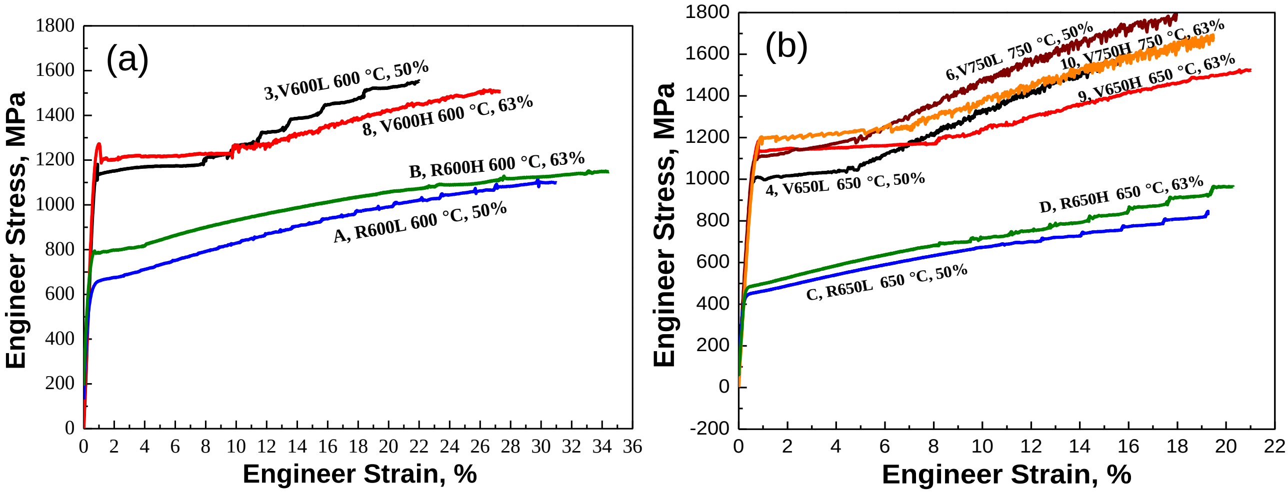
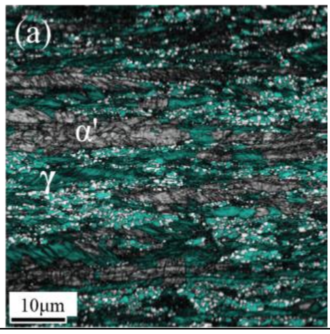
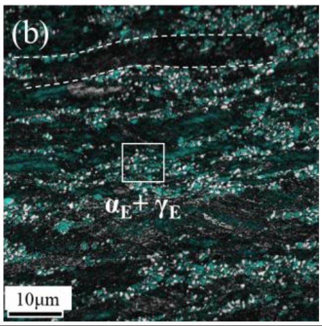
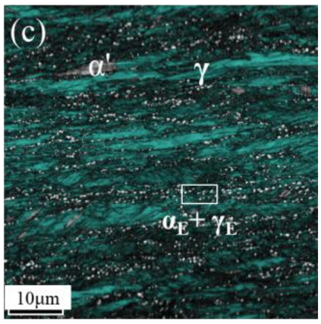


Figure 2. Comparison of engineering stress-strain curves of V-10Mn and 10Mn steels rolled with different reduction ratios at 600 °C (a), at 650 °C and 750 °C (b). The exact warm rolling process listed in Table 2.

thickness positions	Studied specimens		
	V400L	V600L	V750L
Subsurface			

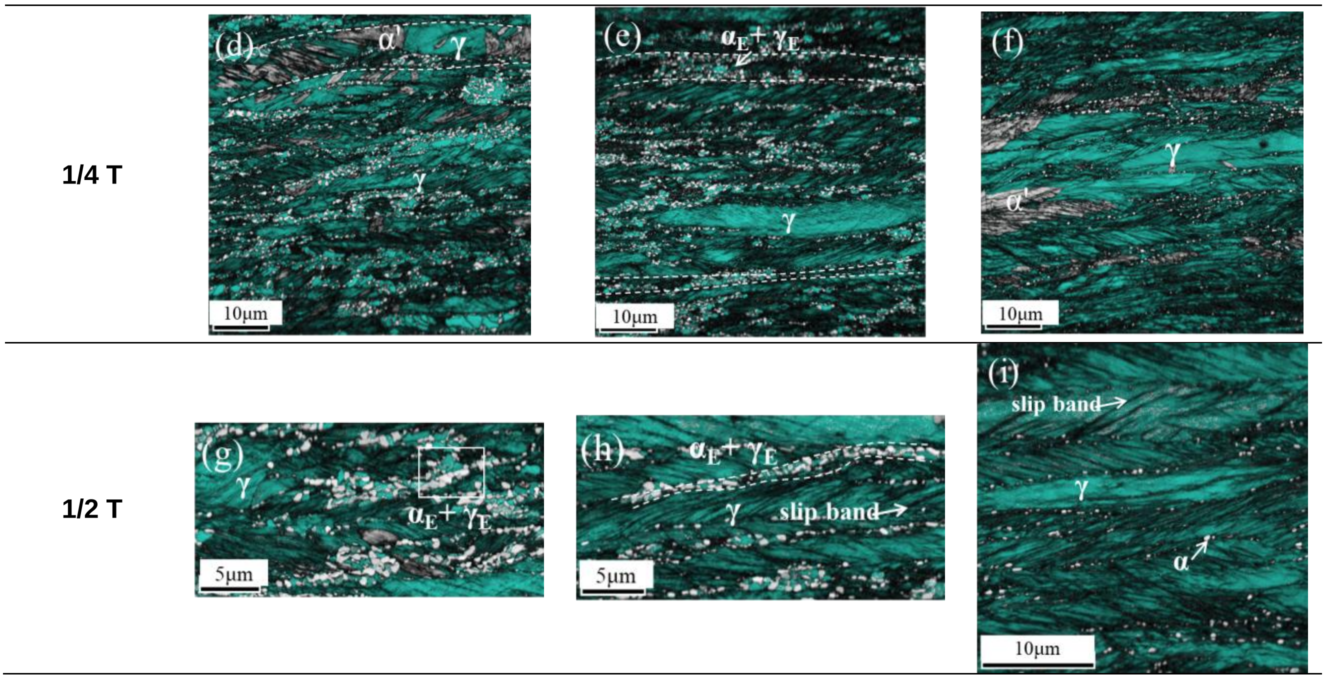


Figure 3 EBSD band contrast images overlapped with phase distribution for microstructures of studied specimens (400L, 600L, 750L) along TD at the different thickness positions. 1/4 T and 1/2 T mean the positions at a quarter and a half of thickness of steel sheet. The blue color denotes FCC phase while the rest BCC phase.

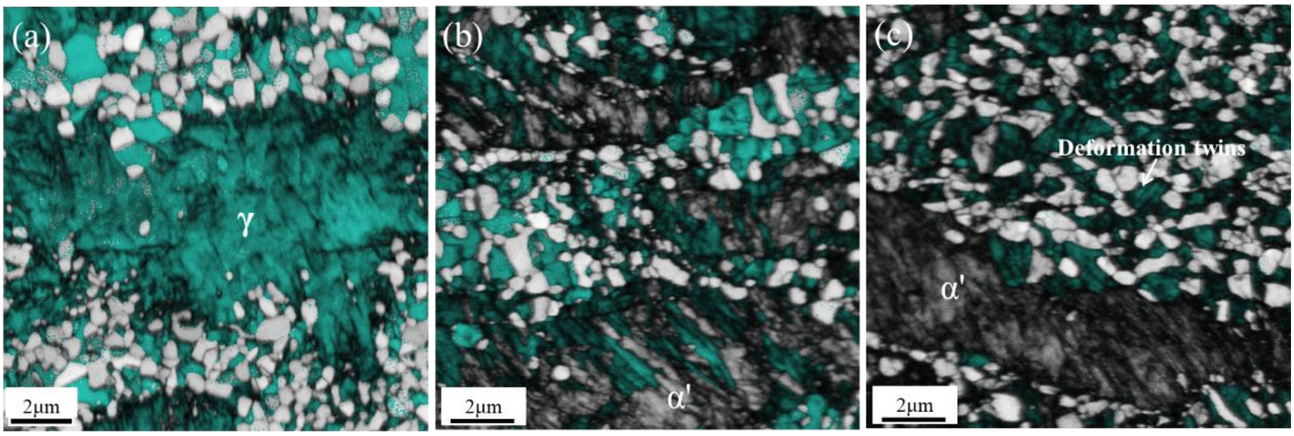
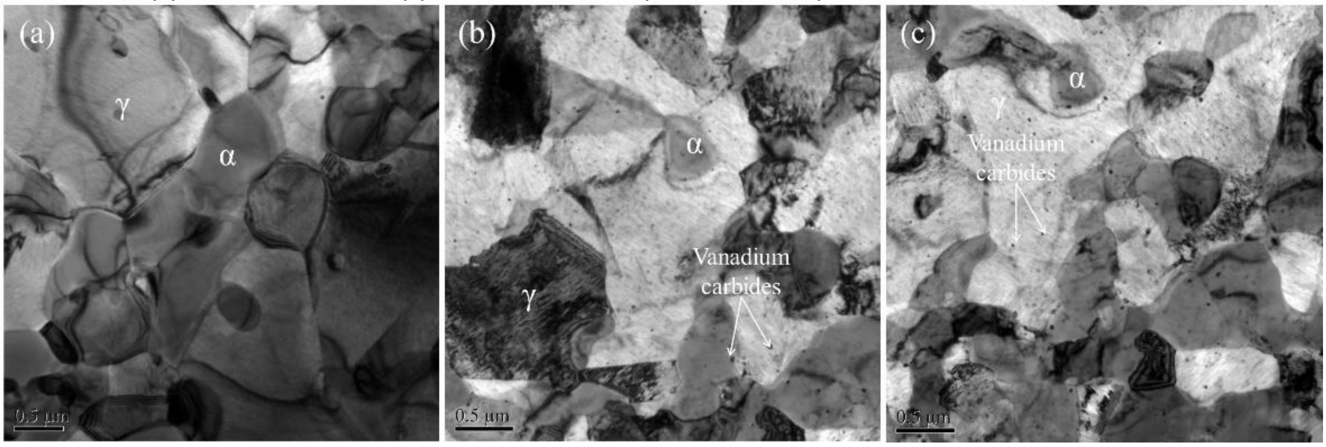


Figure 4 EBSD results on the microstructure evolution of V600H specimen during tensile deformation, examined along the ND at the subsurface. The microstructures before deformation (a), deformed to the true strain of 9% (b) and after fracture (c). The blue color represents FCC phase.



Figue 5 TEM bright field images of microstructures for 10Mn (a) and 10Mn-V steel (b, c), which were warm rolled at 600 ºC by the reduction ratios of 50% (a, b) and 63% (c) and then intercritical annealed at 620 ºC for 5h. The nanosized V-rich carbides extensively precipitated in both ferritic and austenitic phases of 10Mn-V steel but just a few coarse cementite particles in 10Mn steel.

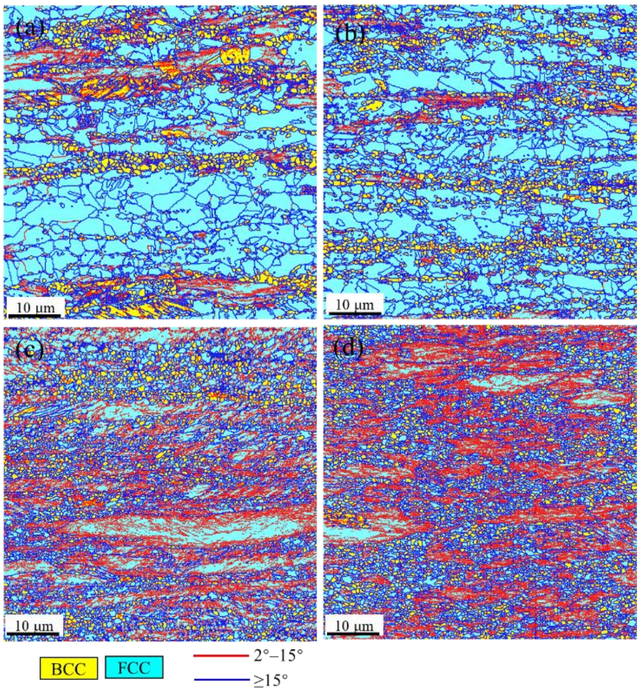


Figure 6 EBSD phase distribution maps for R600L (a), R600H (b), V600L (c) and V600H (d) specimens. The red and blue lines are the low- and high-angle grain boundaries with misorientation higher than 2° and 15°. All microstructures were examined along TD at a quarter of thickness arter and a half of thickness of steel sheet. The blue color denotes FCC phase while the rest BCC phase.

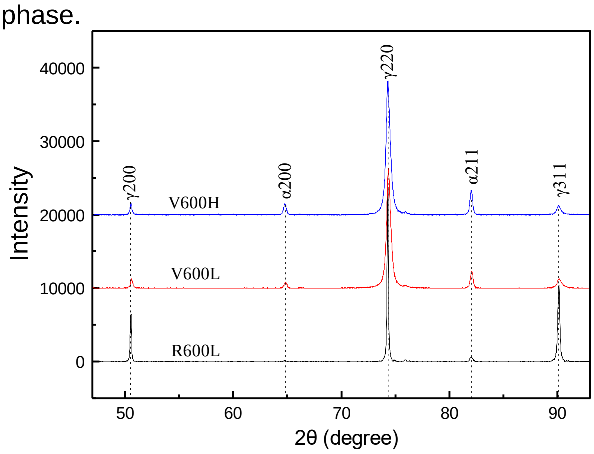


Figure 7 XRD spectra for R600L, V600L and V600H specimens, examined along the ND at a quarter of thickness.

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